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Electrical Properties of Dislocations in Plastically Deformed Float Zone Silicon

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Abstract. — The electrical activity of dislocations created in plastically deformed Float Zone (FZ) silicon wafers have been investigated by means of Deep Levels Transient Spectroscopy (DLTS), of Light Beam Induced Current (LBIC) mappings and of *I-V* Curves. It was found that these dislocations recombine minority carriers due to the generation of deep traps associated to dangling bonds and the aggregation of point defects. Dislocations induce also a soft breakdown in reverse biased aluminium silicon diodes by means of microplasmas. Annealing of the deformed wafers at 1000 °C for 1 hour reduces drastically the deep trap density and suppresses the soft breakdown in Al-Si diodes, probably due to the modification of the point defect atmosphere and the reconstruction of the dislocation core.

1. Introduction

Dislocations are known to strongly influence the electrical properties of electronic devices. Particulary in junctions, this influence can be fairly well evaluated by the mechanism of recombinaison-generation of minority carriers in the depletion region and by the recombination strength in the quasi-neutral region of the bulk. These defects could also be involved in leakage current mechanism. The different possibilities have previously been discussed from a theoretical point of view [1].

The observed effects are correlated with recombination centers related to the dislocations. Thus, it has been established that the electrical activity of dislocations not only depends on the dislocation morphology and on the defects present in the core of the dislocations (intrinsic properties) but also on the presence of metallic impurities and on the oxygen concentration (extrinsic properties). For these reasons, it is necessary to control the dislocation structure and the nature of contaminants. These two conditions are satisfied by the use of plastic deformation of FZ silicon wafers (which minimize the influence of oxygen).

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In addition, when dislocations are formed by plastic deformation, it is well known that, due to dislocation-point defect interaction, their electrical properties can strongly depend on the temperature of deformation (T_d) and also on thermal treatments [2–5]. Such treatments can change the dislocation charge or the dislocation recombination strength in different ways, e.g an increase of the temperature can decrease the dislocation charge and therefore reduces the electrical barrier around these defects [5,6], enhancing their recombination strength [3].

Kimmerling and Patel [7], like Kveder *et al.* [8] already reported that the concentrations of most hole traps analyzed by Deep Level Transient Spectroscopy (DLTS) were reduced markedly after the annealing of a deformed crystal while the density of dislocations was not modified significantly. Ono and Sumino [9] proposed that plastic deformation involves several kinds of hole traps in p-type silicon, denoted by DH(0.24), DH(0.33) and DH(0.56). DH(0.33) was attributed by these authors to jogs and kinks, while the other two were related to agglomerations of point defect resulting from dislocation debris. Kveder *et al.* [8] have obtained three peaks, located at 0.25 eV, 0.39 eV and 0.67 eV above the valence band and they attributed the first level to point defects in the vicinity of dislocations, and the others to dangling bonds.

The aim of the present work is to investigate the electrical activity of dislocations created in oxygen poor silicon wafers by plastic deformation in FZ samples. The investigations involve the characterization of the majority carrier trap levels by DLTS, the evaluation of the recombination strength of the defects by means of Light Beam Induced Current (LBIC) maps, and by the analysis of current-voltage (I-V) Curves of metal semiconductor diodes. Plastically deformed samples are studied in dislocation free regions and in regions containing these defects before and after subsequent annealings.

It is found that the dislocations have a noticeable recombination strength which can be ascribed to the deep levels found in reference [8], that the soft breakdown of aluminium-silicon diodes depends on this strength, and that a subsequent annealing at 1000 °C reduces drastically the influence of these defects.

2. Experimental

Rectangular samples were cut-out from boron-doped ($N_a = 3 \times 10^{14}$ cm^{-3}) FZ (111) silicon wafers. They were optically polished on one face and chemically polished with the C.P.4 etching solution (HF; HNO_3; HCH_3CO_2; 1:8:3) in order to prevent the nucleation of undesired dislocations. Dislocation sources were introduced by scratching the samples in a direction parallel to [110]; then dislocation half-loops were developed by cantilever bending of the samples along the tranversal axis and heating under stress for a period of 1 hour at $T_d = 700$ °C in argon atmosphere.

In this case, in accordance with reference [10], two glide systems were predominant, namely $a/2$ [110](111) and $a/2$ [101](111). Most of dislocations have their Burgers vector parallel to the surface (Fig. 1), their emerging segments were at 60° orientation and homogeneously distributed across the surface. Figure 1 describes schematically the disposition of the dislocations. Notice that a dense array of dislocations, parallel to the surface of the sample is also formed, and is equivalent to a subgrain boundary. The dislocation density varied in the range 5×10^4 cm^{-2} to 10^6 cm^{-2} [10].

Metal semiconductor barriers (Ø= 1.5 mm) were formed by thermal evaporation of 300 nm thick aluminium layer on dislocation free and dislocation containing regions. In the latter case, the emerging segments of dislocations cross the depletion region of the barriers. The metal semiconductor structure was choosen in order to avoid an additional heat treatment needed to realize a $p-n$ junction, and in order to apply subsequent annealings to the samples after the removal of the metallic layer. Indeed, annealings at 1000 °C for 1 hour in an argon flow were

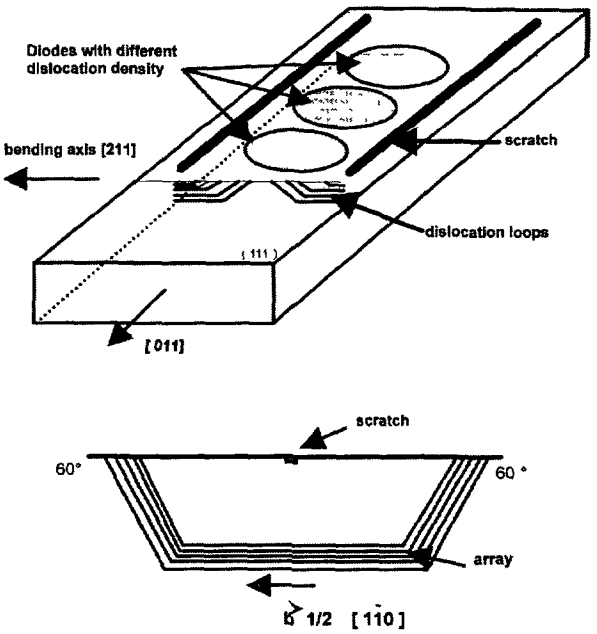

Fig. 1. — Dislocation loops developed from diamond scratches by cantilever bending.

carried out on deformed samples.

Deep Level Transient Spectroscopy measurements were made with a lock-in amplifier technique (P.A.R 410, $I_{\max} = 1$ mA) [11]. To compare the spectra obtained with different samples, the following parameters have been choosen: bias voltage -5 V, filling pulse 5 V with pulse duration 1ms and lock-in frequency 21 Hz.

Light Beam Induced Current (LBIC) mappings were done with a monochromatic light spot (less than 10 μm in diameter) from a monochromator and the focusing system of a metallographic microscope while an $x - y$ stage moved the sample. The diffusion lengths (L_n) were calculated from the spectral variation of the local quantum efficiency in the near infrared range correlated with that of the optical absorption coefficient. Details of these techniques have been already published [12].

I-V Curves were obtained using the Keithley 237 Source Measurement Unit when the diodes were reverse biased.

Finally, the dislocation density N_{dis} was measured by counting the dislocation etch pits after selective chemical etching (Sirtl etch).

3. Results

3.1. As-Received Material. — LBIC mappings show no variation of photocurrent response and the initial diffusion length was $L_n = 150$ μm. The absence of recombination centers in the control material is confirmed by DLTS spectra in wich no signal is detected. The reverse *I-V* characteristic of these dislocation free diodes shows that the reverse current is independent of applied voltage and no soft-breakdown is observed for voltages below 10 V.

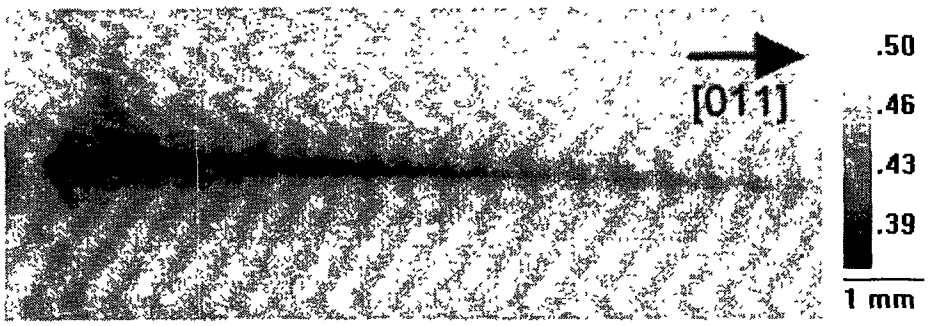

Fig. 2. — L.B.I.C photocurrent map of a deformed sample (arbitrary units).

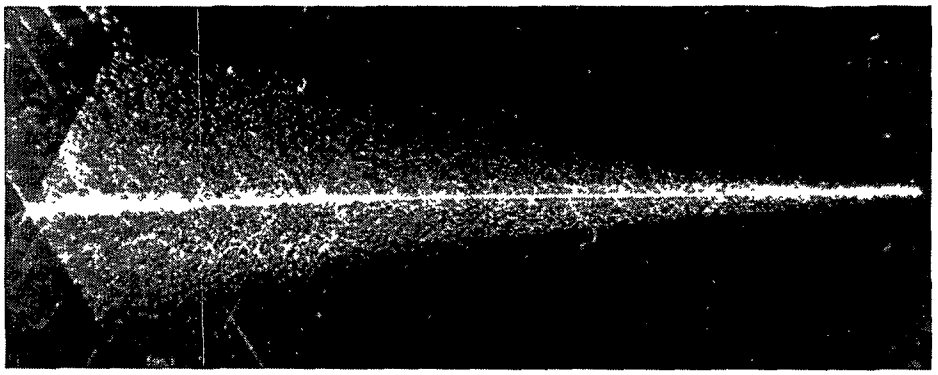

Fig. 3. — Optical microphotography of the etch-pit distribution corresponding to Figure 2.

3.2. Dislocated Material. — After deformation at $T_d = 700$ °C for a period of 1 hour, dislocations are formed and the samples are first characterized by LBIC technique. The LBIC mapp given in Figure 2 illustrates the spatial distribution of the defects induced by plastic deformation and their recombination strength. As shown in Figure 3, the distribution of etch pits revealed by Sirtl etch (confirmed by X-ray topography), corresponds to the attenuation of the photocurrent in the mapp of Figure 2. Thus, as expected [13], the dislocations decrease the photocurrent and hence the diffusion length (L_n). The value of L_n is found to decrease with the dislocation density, up to $L_n = 20$ μm, when $N_{dis} = 10^6$ cm^{-2}. The minority carrier diffusion length mapp of Figure 4 illustrates this dependency. This mapp was obtained on a sample in which several scratches were used to increase the dislocation density.

Notice that, after the deformation, the photocurrent response was unchanged in dislocation free regions.

The DLTS spectra carried out on dislocated regions are shown in Figure 5 and Figure 6. We verified that the DLTS signal increases with the dislocations density. Three broads peaks are distinguished in the deformed samples before annealing and their characteristics are in agreement with those observed in previous works [7–9].

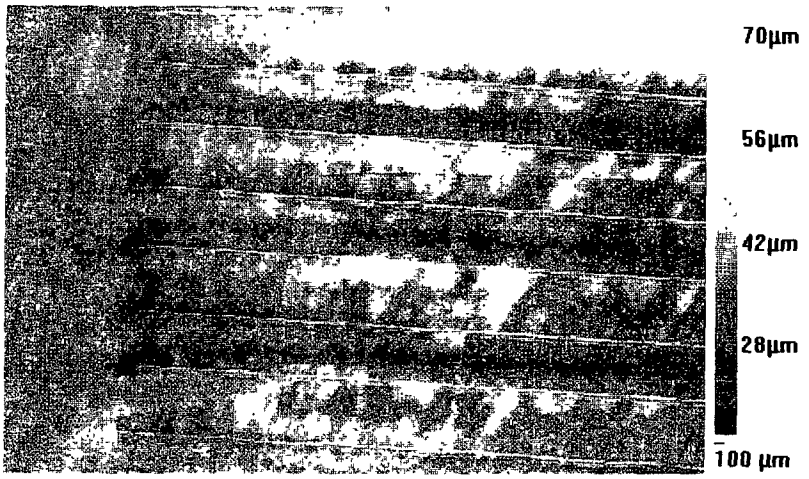

Fig. 4. — Diffusion length map computed thanks to several L.B.I.C scannings at different wavelengths. This FZ sample, oriented < 111 >, contains several scratches.

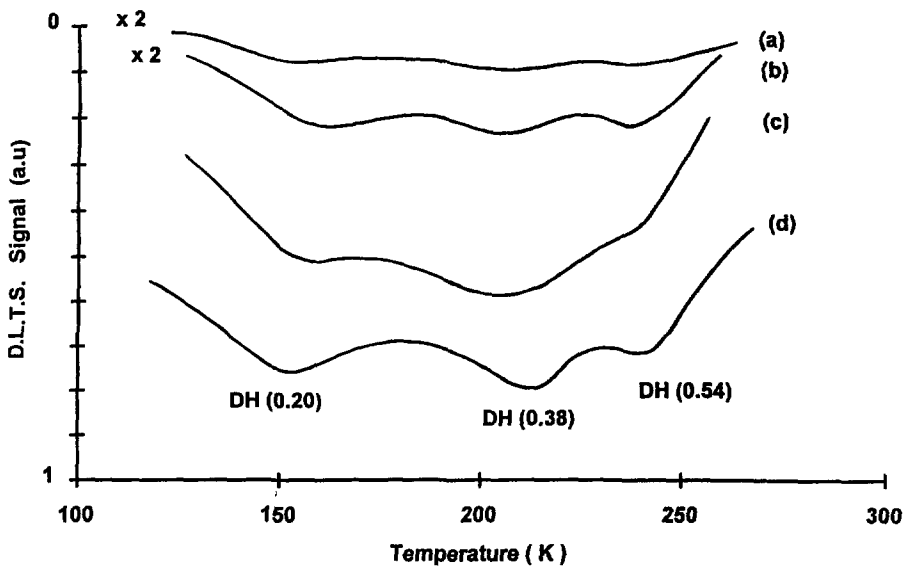

Fig. 5. — D.L.T.S spectra of deformed p-type silicon (bias voltage -5 V, filling pulse 5 V with pulse duration 1 ms and lock-in frequency 21 Hz). a) $N_{dis} = 5 \times 10^4$ cm^{-2} b) $N_{dis} = 10^5$ cm^{-2} c) $N_{dis} = 6 \times 10^5$ cm^{-2} d) $N_{dis} = 9 \times 10^5$ cm^{-2}

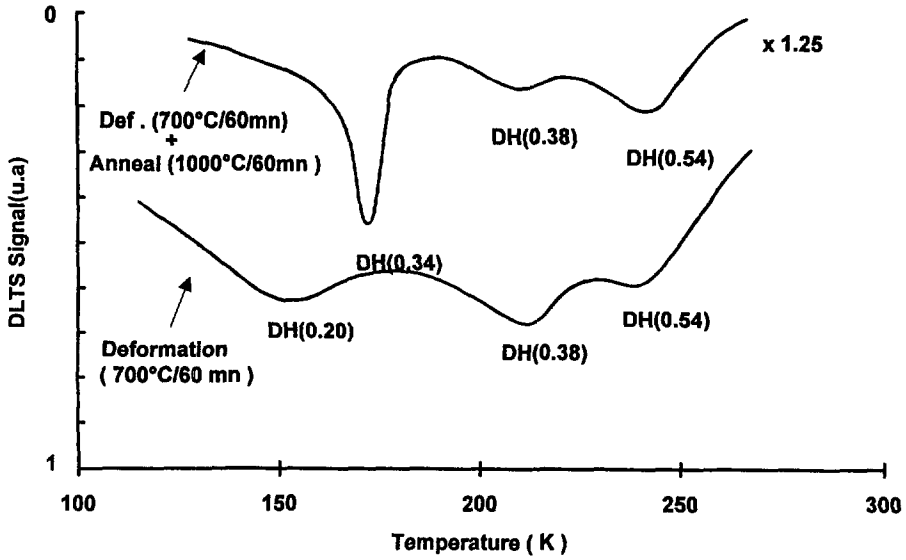

Fig. 6. — D.L.T.S spectra of deformed p-type silicon before and after annealing at 1000 °C for 1 hour (bias voltage −5 V, filling pulse 5 V with pulse duration 1 ms and lock-in frequency 21 Hz).

Table I. — *D.L.T.S results. Concentration of holes traps (N_t) and energy levels (E_t) detected in deformed samples before and after annealing at 1000 °C for 1 hour in argon.*

	DH (0.20)	DH (0.34)	DH (0.38)	DH (0.54)
	$E_t = 0.20$ eV	$E_t = 0.34$ eV	$E_t = 0.38$ eV	$E_t = 0.54$ eV
Deformation 700 °C/60 mn	$N_t = 8.2\times10^{12}$ cm^{-3}		$N_t = 9\times10^{12}$ cm^{-3}	$N_t = 8\times10^{12}$ cm^{-3}
Deformation 700 °C/60 mn + Anneal 1000 °C60 mn		$N_t = 5.5\times10^{12}$ cm^{-3}	$N_t = 2\times10^{12}$ cm^{-3}	$N_t = 3\times10^{12}$ cm^{-3}

The hole traps related to these broad peaks are labeled DH(0.20), DH(0.38) and DH(0.54), respectively, with their energy levels given in eV in the parentheses. The concentration of these traps N_t is evaluated from the magnitude of each peak of the spectrum and listed in Table I.

The reverse I-V Curves of Schottky barriers formed on dislocated crystal are presented in Figure 7. A soft breakdown is observed which does not appear in the dislocation free diodes. Althought a heat treatment is needed for the deformation, the diodes made in the dislocation free region have I-V Curves similar to those obtained in as received samples. Thus, the breakdown can be attributed to the dislocations and not to the thermal treatment. In the soft-breakdown the current intensity is found to be dependent on the dislocation density.

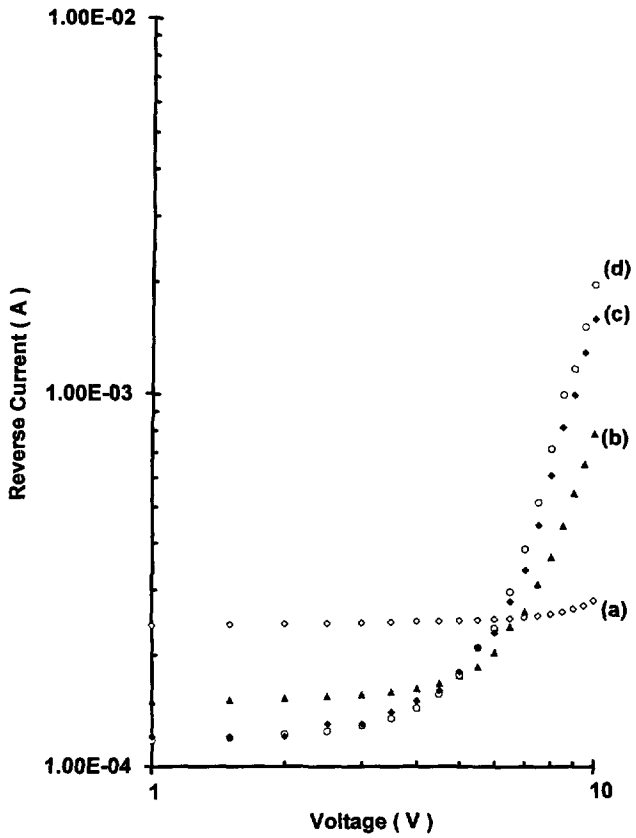

Fig. 7. — Reverses I-V Curves of Schottky barriers formed on dislocated crystal. a) Dislocation free diode. b) $N_{\text{dis}} = 10^5$ cm^{-2} c) $N_{\text{dis}} = 6 \times 10^5$ cm^{-2} d) $N_{\text{dis}} = 9 \times 10^5$ cm^{-2}

3.3. Dislocated Material After Annealing at 1000 °C. — Dislocated samples were annealed without any stress at 1000 °C for a period of 1 hour in an argon atmosphere.

After this treatment, no appreciable changes were detected in either the density or the distribution of dislocation etch pits.

The DLTS spectra obtained after annealing are also shown in Figure 6. In agreement with previous studies [7, 8], all the peaks which are characteristic of deformed specimen are found to decrease. Table I displays the evolution of N_t for the different peaks. The concentration of majority carrier trap levels is approximately divided by three for DH(0.38) and DH(0.54) while N_t related to DH(0.20) could not be calculated. In fact, this peak is not detected due to the presence of a fourth peak, with any activation energy of about 0.34 eV, and which is revealed in our moderately dislocated crystal ($N_{\text{dis}} = 8 \times 10^5$ cm^{-2}) after annealing at 1000 °C. It is probably due to a contamination during such annealing because this level is also detected in dislocation free diodes.

The reverse I-V Curves obtained on the same dislocated sample before and after annealing are given in Figure 8. After annealing, a decrease of the reverse current is observed while the soft-breakdown has disappeared. As shown in Figure 9, the current measured in defect

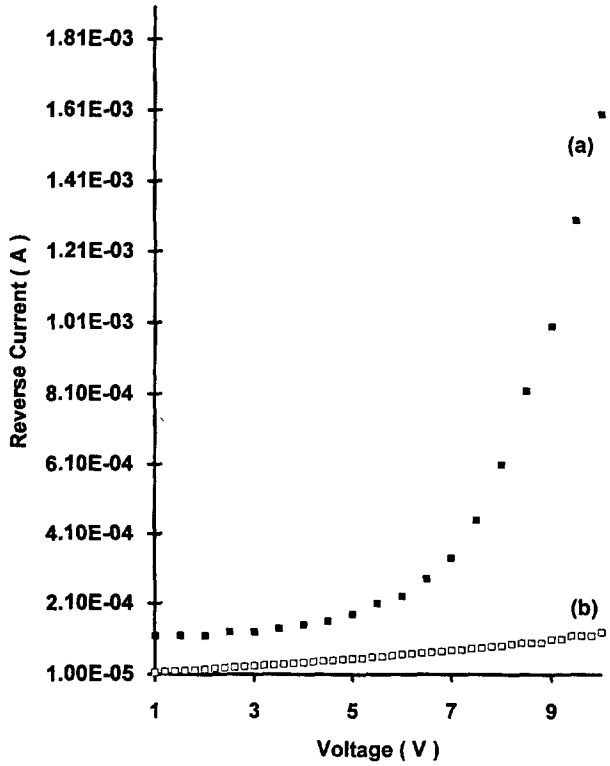

Fig. 8. — Reverse I-V Curves of dislocation containing diodes ($N_{dis} = 9 \times 10^5$ cm^{-2}). a) After deformation. b) After deformation and subsequent annealing at 1000 °C for 1 hour.

containing diodes is similar to that observed in dislocation free diodes. Note that the influence of a high temperature thermal treatment on dislocation free diodes decrease by a factor two the reverse current for low voltage.

4. Discussion and Conclusion

The dislocations created by plastic deformation in FZ silicon wafers are sources of recombination centers. As the oxygen concentration is in 10^{17} cm^{-3} range and the material of high purity, these centers are generated by the defects and not by a contamination during the treatments applied to the wafers.

Our DLTS measurements revealed three levels which are close to those found by Kveder *et al.* [8], and by Ono and Sumino [9]. As proposed in reference [8], it is possible to relate our three levels to clouds of point defects around the dislocations (DH(0.20)) and to states localized in the dislocation core like those due to dangling bonds (DH(0.38) and DH(0.54)). The decrease of the trap density which results from the decrease of the DLTS peak in Figure 5 after the subsequent annealing at 1000 °C, suggests that the point defects atmosphere surrounding the

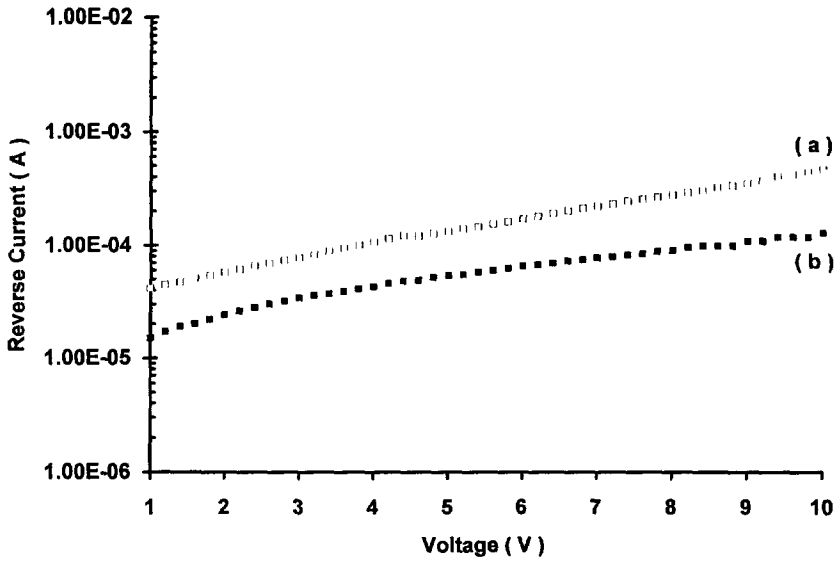

Fig. 9. — Reverses I-V Curves of Schottky barriers after annealing at 1000 °C for 1 hour. a) Dislocation free diode. b) Dislocation containing diode ($N_{\text{dis}} = 9 \times 10^5$ cm^{-2}).

dislocations is modified and it is not preclude that a reconstruction of dangling bonds can occur.

The soft breakdown observed in Figure 7 in the deformed samples for reverse voltages of about 10 V, can be with a large mesure of certainty due to the formation of microplasmas at the emergences of dislocation lines in the space charge region, as the breakdown current increases with the defect density. This breakdown dissapears after the annealing at 1000 °C (Fig. 8), as a consequence of the modification of the dislocation space charge tube. This could be explained by the reconstruction of the dislocation cores or/and by a relaxation of impurities from the dislocation inducing the decrease of electrical activity of the point defect aggregation.

In conclusion, we have shown that dislocations created by plastic deformation in FZ silicon wafers have a noticeable electrical activity, due to the generation of different types of recombination centers, which could be ascribed to dangling bonds and point defects aggregation. These dislocations are able to develop microplasmas when they cross the depletion region of a junction, inducing a soft breakdown in the reverse I-V Curves.

When the deformed wafers are annealed at 1000 °C for 1 hour, the density of trap levels decreases and the soft breakdown disappears, probably due to a modification of the point defects atmosphere and to the reconstruction of the dislocation core.

Acknowledgments

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